(12)

(21) 2 429 439

(22) 23.05.2003

(51) Int. Cl.⁷: **C22C 38/12, C22C 38/04, C22C 38/08, C21D 7/13, C22C 38/20**

(30)   2002-152379   JP   27.05.2002
        2002-377829   JP   26.12.2002

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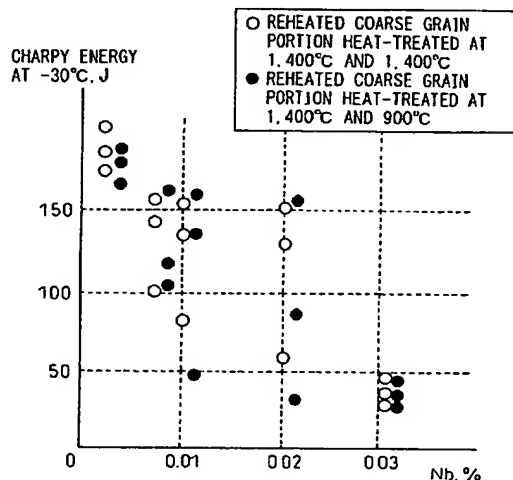
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(54) ACIER A HAUTE RESISTANCE AYANT UNE EXCELLENTE TENACITE A BASSE TEMPERATURE ET DANS DES ZONES AYANT ETE CHAUFFEES LORS DE SOUDURE, METHODE POUR LE PRODUIRE ET METHODE POUR PRODUIREDES TUBES EN CET ACIER

(54) HIGH-STRENGTH STEEL EXCELLENT IN LOW TEMPERATURE TOUGHNESS AND TOUGHNESS AT WELD HEAT-AFFECTED ZONE, METHOD FOR PRODUCING THE SAME, AND METHOD FOR PRODUCING HIGH-STRENGTH STEEL PIPE

(57)   The present invention provides an ultra-high-strength steel pipe excellent in weldability on site and a method for producing the steel pipe by improving the reliability of the low temperature toughness of a steel to which elements to enhance hardenability are added for furthering high-strengthening and also improving toughness at a weld heat affected zone subjected to double or more layer welding and, in the method, the steel is made to consist of a structure composed of bainite and/or martensite by containing prescribed amounts of C, Si, Mn, P, S, Ni, Mo, Nb, Ti, Al and N, and, as occasion demands, one or more of B, V, Cu, Cr,   Ca, REM and Mg, and regulating C, Si, Mn, Cr, Ni, Cu, V and Mo, those being elements to enhance hardenability, by a specific relational expression. The diameter of prior austenite grains may be regulated in a prescribed range.   The method includes the steps of heating a casting to a temperature not lower than the Ac3 point, hot rolling it, and thereafter cooling the resulting hot-rolled steel plate at a prescribed cooling rate.

(54) Titre : ACIER A HAUTE RESISTANCE AYANT UNE EXCELLENTE TENACITE A BASSE TEMPERATURE ET
DANS DES ZONES AYANT ETE CHAUFFEES LORS DE SOUDURE, METHODE POUR LE PRODUIRE ET
METHODE POUR PRODUIRE DES TUBES EN CET ACIER

(54) Title: HIGH-STRENGTH STEEL EXCELLENT IN LOW TEMPERATURE TOUGHNESS AND TOUGHNESS AT
WELD HEAT-AFFECTED ZONE, METHOD FOR PRODUCING THE SAME, AND METHOD FOR PRODUCING
HIGH-STRENGTH STEEL PIPE
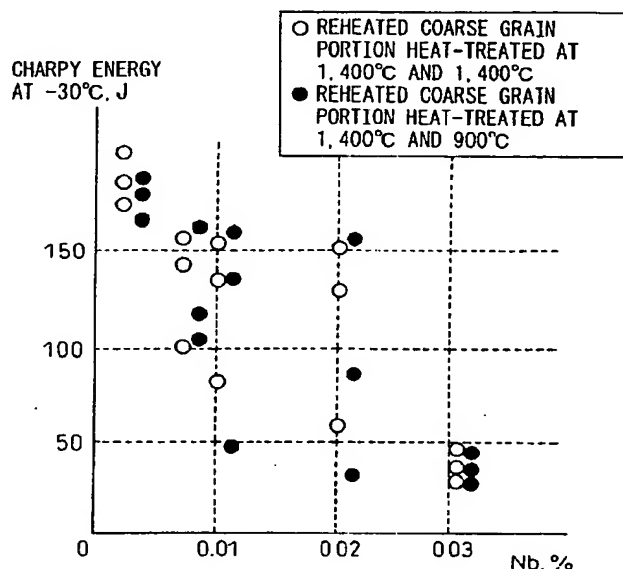
(57) Abrégé/Abstract:
The present invention provides an ultra-high- strength steel pipe excellent in weldability on site and a method for producing the
steel pipe by improving the reliability of the low temperature toughness of a steel to which elements to enhance hardenability are
added for furthering high-strengthening and also improving toughness at a weld heat affected zone subjected to double or more

(57) Abrégé(suite)/Abstract(continued):
layer welding and, in the method, the steel is made to consist of a structure composed of bainite and/or martensite by containing prescribed amounts of C, Si, Mn, P. S, Ni, Mo, Nb, Ti. Al and N, and, as occasion demands, one or more of B, V, Cu, Cr. Ca, REM and Mg, and regulating C, Si, Mn. Cr. Ni, Cu, V and Mo, those being elements to enhance hardenability, by a specific relational expression. The diameter of prior austenite grains may be regulated in a prescribed range. The method includes the steps of heating a casting to a temperature not lower than the Ac_3 point, hot rolling it, and thereafter cooling the resulting hot-rolled steel plate at a prescribed cooling rate.

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<u>HIGH-STRENGTH STEEL EXCELLENT IN LOW TEMPERATURE</u>
<u>TOUGHNESS AND TOUGHNESS AT WELD HEAT-AFFECTED ZONE,</u>
<u>METHOD FOR PRODUCING THE SAME,</u>
<u>AND METHOD FOR PRODUCING HIGH-STRENGTH STEEL PIPE</u>

5

BACKGROUND OF THE INVENTION
1.   Field of the Invention
The present invention relates to methods for
10   producing a ultra-high-strength hot-rolled steel having a
tensile strength of not lower than 800 MPa, in particular
not lower than 900 MPa, and being excellent in toughness
of a base steel and toughness at a weld heat-affected
zone in the temperature range from -60°C to 0°C
15   (hereunder referred to as "low temperature toughness" and
"weld heat-affected zone toughness"), and for producing a
steel plate and a steel pipe made of the hot-rolled
steel.

Such ultra-high-strength hot-rolled steels are,
20   after being further processed and welded, widely used for
line pipes for the transport of natural gas or crude oil,
pressure vessels, welded structures and the like.
2.   Description of the Related Art
In recent years, a steel plate for a line pipe,
25   for water pumping (for a penstock for example), or for a
pressure vessel is required to have improved high
strength and low temperature toughness. For example, in
the case of a steel plate for a line pipe, various
studies have already been undertaken with regard to the
30   production of a ultra-high-strength steel plate having a
tensile strength of not lower than 800 MPa (not lower
than X100 in the API standard) and high-strength steels
excellent in low temperature toughness, weld heat-
affected zone toughness and weldability are disclosed in
35 · Japanese Patent Nos. 3244986 and 3262972. In addition, a
ultra-high-strength line pipe having a tensile strength
of not lower than 900 MPa and the production method

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thereof are disclosed in Japanese Unexamined Patent
Publication No. 2000-199036.

However, in a steel plate for a line pipe
disclosed in the above-mentioned Japanese Patent Nos.
3244986 and 3262972, though the Charpy absorbed energy at
-20°C at a heat-affected zone to which single layer
welding is applied is not lower than 100 J and thus very
good, weld heat-affected zone toughness sometimes
deteriorates at a heat-affected zone to which double or
more layer welding is applied under some welding
conditions.

Further, in a steel plate for a line pipe
disclosed in the above-mentioned Japanese Patent Nos.
3244986 and 3262972 and also in a ultra-high-strength
line pipe disclosed in the above-mentioned Japanese
Unexamined Patent Publication No. 2000-199036, though the
Charpy absorbed energy of a base steel at -40°C is not
lower than 200 J on the average when the number of the
specimens (hereunder referred to as "n") subjected to the
test by using the same material and under the same test
conditions is three and the result is very good, the
problem here is that the Charpy absorbed energy of some
specimens is lower than 200 J and is dispersed widely in
some cases.

As a result of studying the problem of the
dispersion of low temperature toughness in detail, it was
clarified that Charpy absorbed energy was lower than
about 200 J with a probability of about 20 percent when
Charpy impact test was performed at -40°C under an
increased number n, and further that Charpy absorbed
energy of some specimens was not higher than 100 J and
brittle fractured faces were observed on the fractured
surfaces of the specimens when the Charpy impact test was
performed in the temperature range from -60°C to not
higher than -40°C.

Meanwhile, the present inventors proposed a
method for improving low temperature toughness by

contriving a welding method described in Japanese Patent·
Application No. 2001-336670. However, it was also
clarified that the proposed method was not immediately
applicable because it was not suitable for mass
5   production and required the introduction of new
equipment. In view of the above situation, the
development of a high-strength line pipe excellent in low
temperature toughness at both a base steel and a weld is
required.

10   SUMMARY OF THE INVENTION

The present invention provides a ultra-high-strength
steel having a tensile strength of not lower than 800 MPa
and a steel pipe made thereof, the steel being excellent
in weld heat-affected zone toughness, particularly in
15   shelf energy at a weld heat-affected zone when multi-
layer welding is applied; having a Charpy absorbed energy
of a base steel at -40°C being not lower than 200 J on
the average and with little dispersion; having excellent
low temperature toughness; and further being easily
20   weldable at a site. Here, shelf energy is Charpy
absorbed energy measured in the temperature range where a
material ductilely fractures at one hundred percent when
a Charpy impact test is applied at various temperatures
to the material that brittlely fractures at a low
25   temperature.

The present inventors carried out intensive studies
on the chemical components of a steel material and the
microstructure thereof for obtaining a high-strength
steel having a tensile strength of not lower than 800 MPa
30   (not lower than X100 in the API standard); having shelf
energy of not lower than 100J at a weld heat-affected
zone to which multi-layer welding is applied;
having Charpy absorbed energy of a base steel not lower
than 200 J on the average and with little dispersion in
35   the temperature range of not higher than -40°C; and
further being easily weldable on site.

As a result of the studies, firstly, the present

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inventors made it clear that the deterioration of low
temperature toughness in double layer welding was caused
by Nb carbonitride and then confirmed that the reduction
of an Nb amount was extremely effective in avoiding the
5   deterioration. Secondly, with regard to a base steel,
low Charpy absorbed energy was observed sometimes under
some test conditions, and the present inventors made it
clear that the low Charpy absorbed energy was caused by
coarse grains which are partially existing, and found
10   that the reduction of an Nb amount was extremely
effective as a countermeasure.
         The present invention of a high-strength steel
excellent in low temperature toughness and weld heat-
affected zone toughness was accomplished by further
15   controlling a P value that was an index of hardenability
in an appropriate range for enhancing strength that was
lowered once by the decrease in an Nb amount.
         The present invention was established on the basis
of the above findings and the gist thereof is as follows:
20   (1) A high-strength steel excellent in low
temperature toughness and weld heat-affected zone
toughness, characterized by: containing, in mass,
             C: 0.02 to 0.10%,
             Si: not more than 0.6%,
25   Mn: 1.5 to 2.5%,
             P: not more than 0.015%,
             S: not more than 0.003%,
             Ni: 0.01 to 2.0%,
             Mo: 0.2 to 0.6%,
30   Nb: less than 0.010%,
             Ti: not more than 0.030%,
             Al: not more than 0.070%, and
             N: not more than 0.0060%,
with the balance consisting of Fe and unavoidable
35   impurities; the P value of the steel defined by the
following expression being in the range from 1.9 to 3.5;
and the microstructure of the steel being mainly composed

of martensite and bainite:

$P = 2.7C + 0.4Si + Mn + 0.8Cr + 0.45(Ni + Cu) + 2V + Mo - 0.5$.

(2) A high-strength steel excellent in low temperature toughness and weld heat-affected zone toughness, characterized by: containing, in mass,

C: 0.02 to 0.10%,

Si: not more than 0.6%,

Mn: 1.5 to 2.5%,

P: not more than 0.015%,

S: not more than 0.003%,

Ni: 0.01 to 2.0%,

Mo: 0.1 to 0.6%,

Nb: less than 0.010%,

Ti: not more than 0.030%,

B: 0.0003 to 0.0030%,

Al: not more than 0.070%, and

N: not more than 0.0060%, so as to satisfy the expression $Ti - 3.4N \geqq 0$,

with the balance consisting of Fe and unavoidable impurities; the P value of the steel defined by the following expression being in the range from 2.5 to 4.0; and the microstructure of the steel being composed of martensite and bainite:

$P = 2.7C + 0.4Si + Mn + 0.8Cr + 0.45(Ni + Cu) + 2V + 1.5Mo$.

(3) A high-strength steel excellent in low temperature toughness and weld heat-affected zone toughness according to the item (1) or (2), characterized by further containing, in mass, one or more of

V: 0.001 to 0.10%,

Cu: 0.01 to 1.0%,

Cr: 0.01 to 1.0%,

Ca: 0.0001 to 0.01%,

REM: 0.0001 to 0.02%, and

Mg: 0.0001 to 0.006%.

(4) A high-strength steel excellent in low

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temperature toughness and weld heat-affected zone
toughness according to any one of the items (1) to (3),
characterized by the average diameter of the prior
austenite grains in the steel being not larger than 10

5   μm.

(5) A high-strength steel excellent in low
temperature toughness and weld heat-affected zone
toughness, characterized by: containing, in mass,

C: 0.02 to less than 0.05%,

10   Si: not more than 0.6%,

Mn: 1.5 to 2.5%,

P: not more than 0.015%,

S: not more than 0.001%,

Ni: 0.01 to 2.0%,

15   Mo: 0.1 to 0.6%,

Nb: less than 0.010%,

Ti: not more than 0.030%,

B: 0.0003 to 0.0030%,

Al: not more than 0.070%, and

20   N: not more than 0.0060%, so as to satisfy the
expression $Ti - 3.4N \geq 0$, and further one or more of

V: 0.001 to 0.10%,

Cu: 0.01 to 1.0%, and

Cr: 0.01 to 1.0%,

25   with the balance consisting of Fe and unavoidable
impurities; the P value of the steel defined by the
following expression being in the range from 2.5 to 4.0;
the microstructure of the steel being composed of
martensite and bainite; and the average diameter of the

30   prior austenite grains in the steel being not larger than
10 μm:

$P = 2.7C + 0.4Si + Mn + 0.8Cr + 0.45(Ni + Cu) + 2V + 1.5Mo$.

(6) A high-strength steel excellent in low

35   temperature toughness and weld heat-affected zone
toughness, characterized by: containing, in mass,

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C: 0.02 to less than 0.05%,

Si: not more than 0.6%,

Mn: 1.5 to 2.5%,

P: not more than 0.015%,

5   S: not more than 0.001%,

Ni: 0.01 to 2.0%,

Mo: 0.1 to 0.6%,

Nb: less than 0.010%,

Ti: not more than 0.030%,

10   B: 0.0003 to 0.0030%,

Al: not more than 0.070%, and

N: not more than 0.0060%, so as to satisfy the

expression $Ti - 3.4N \geq 0$, and further one or more of

V: 0.001 to 0.10%,

15   Cu: 0.01 to 1.0%,

Cr: 0.01 to 1.0%, and

Ca: 0.0001 to 0.01%,

with the balance consisting of Fe and unavoidable
impurities; the P value of the steel defined by the

20   following expression being in the range from 2.5 to 4.0;
the microstructure of the steel being composed of
martensite and bainite; and the average diameter of the
prior austenite grains in the steel being not larger than
10 μm:

25         $P = 2.7C + 0.4Si + Mn + 0.8Cr + 0.45(Ni + Cu) + 2V + 1.5Mo$.

(7) A method for producing a high-strength steel
plate excellent in low temperature toughness and weld
heat-affected zone toughness, the method being the one

30   for producing a steel plate from a casting containing
components according to any one of the items (1) to (3),
(5) and (6), characterized by: reheating the casting to a
temperature of not lower than the Ac_3 point; hot rolling
it; and thereafter cooling the resulting steel sheet at a

35   cooling rate not lower than 1°C/sec. to a temperature not
higher than 550°C.

(8) A method for producing a high-strength steel pipe excellent in low temperature toughness and weld heat-affected zone toughness according to the item (7), characterized by: cold-forming a cooled steel plate into a pipe; and thereafter applying seam welding to the abutted portion thereof.

(9) A high-strength steel pipe excellent in low temperature toughness and weld heat-affected zone toughness, characterized by, in the pipe having a seam-welded portion: the base steel containing, in mass,

    C: 0.02 to 0.1%,
    Si: not more than 0.8%,
    Mn: 1.5 to 2.5%,
    P: not more than 0.015%,
    S: not more than 0.003%,
    Ni: 0.01 to 2%,
    Mo: 0.2 to 0.8%,
    Nb: less than 0.010%,
    Ti: not more than 0.03%,
    Al: not more than 0.1%, and
    N: not more than 0.008%,

with the balance consisting of Fe and unavoidable impurities; the P value defined by the following expression being in the range from 1.9 to 4.0; and the microstructure being mainly composed of martensite and bainite:

$$P = 2.7C + 0.4Si + Mn + 0.8Cr + 0.45(Ni + Cu) + 2V + Mo - 0.5.$$

(10) A high-strength steel pipe excellent in low temperature toughness and weld heat-affected zone toughness, characterized by, in the pipe having a seam-welded portion: the base steel containing, in mass,

    C: 0.02 to 0.10%,
    Si: not more than 0.8%,
    Mn: 1.5 to 2.5%,
    P: not more than 0.015%,
    S: not more than 0.003%,

Ni: 0.01 to 2%,

Mo: 0.1 to 0.8%,

Nb: less than 0.010%,

Ti: not more than 0.030%,

B: 0.0003 to 0.003%,

Al: not more than 0.1%, and

N: not more than 0.008%, so as to satisfy the
expression $Ti - 3.4N \geqq 0$,
with the balance consisting of Fe and unavoidable
impurities; the P value defined by the following
expression being in the range from 2.5 to 4.0; and the
microstructure being mainly composed of martensite and
bainite:

$P = 2.7C + 0.4Si + Mn + 0.8Cr + 0.45(Ni + Cu) + 2V + 1.5Mo$.

(11) A high-strength steel pipe excellent in low
temperature toughness and weld heat-affected zone
toughness according to the item (9) or (10),
characterized by further containing, in mass, one or more
of

V: 0.001 to 0.3%,

Cu: 0.01 to 1%,

Cr: 0.01 to 1%,

Ca: 0.0001 to 0.01%,

REM: 0.0001 to 0.02%, and

Mg: 0.0001 to 0.006%.

(12) A high-strength steel pipe excellent in low
temperature toughness and weld heat-affected zone
toughness according to any one of the items (9) to (11),
characterized by the average diameter of the austenite
grains in the steel pipe being not larger than 10 μm.

(13) A high-strength steel pipe excellent in low
temperature toughness and weld heat-affected zone
toughness, characterized by, in the pipe having a seam-
welded portion: the base steel containing, in mass,

C: 0.02 to less than 0.05%,

Si: not more than 0.8%,

Mn: 1.5 to 2.5%,

P: not more than 0.015%,

S: not more than 0.001%,

5   Ni: 0.01 to 2%,

Mo: 0.1 to 0.8%,

Nb: less than 0.010%,

Ti: not more than 0.030%,

B: 0.0003 to 0.003%,

10   Al: not more than 0.1%, and

N: not more than 0.008%, so as to satisfy the

expression Ti - 3.4N ≧ 0, and further one or more of

V: 0.001 to 0.3%,

Cu: 0.01 to 1%, and

15   Cr: 0.01 to 1%,

with the balance consisting of Fe and unavoidable
impurities; the P value defined by the following
expression being in the range from 2.5 to 4.0; the
microstructure being mainly composed of martensite and
20   bainite; and the average diameter of the austenite grains
being not larger than 10 μm:

$$P = 2.7C + 0.4Si + Mn + 0.8Cr + 0.45(Ni + Cu) + 2V + 1.5Mo.$$

(14) A high-strength steel pipe excellent in low
25   temperature toughness and weld heat-affected zone
toughness, characterized by, in the pipe having a seam-
welded portion: the base steel containing, in mass,

C: 0.02 to less than 0.05%,

Si: not more than 0.8%,

30   Mn: 1.5 to 2.5%,

P: not more than 0.015%,

S: not more than 0.003%,

Ni: 0.01 to 2%,

Mo: 0.1 to 0.8%,

35   Nb: less than 0.010%,

Ti: not more than 0.030%,

B: 0.0003 to 0.003%,

Al: not more than 0.1%, and

N: not more than 0.008%, so as to satisfy the
expression Ti - 3.4N $\geq$ 0, and further one or more of

5   V: 0.001 to 0.3%,

Cu: 0.01 to 1%,

Cr: 0.01 to 1%, and

Ca: 0.0001 to 0.01%,
with the balance consisting of Fe and unavoidable

10   impurities; the P value defined by the following
expression being in the range from 2.5 to 4.0; the
microstructure being mainly composed of martensite and
bainite; and the average diameter of the austenite grains
being not larger than 10 μm:

15      $P = 2.7C + 0.4Si + Mn + 0.8Cr + 0.45(Ni + Cu) + 2V +$
$1.5Mo.$

(15) A method for producing a high-strength steel
pipe excellent in low temperature toughness and weld
heat-affected zone toughness, characterized by: reheating

20   the casting containing components according to any one of
the items (9) to (14) to a temperature of not lower than
the Ac_3 point; hot rolling it; thereafter cooling the
resulting steel sheet at a cooling rate not lower than
1°C/sec. to a temperature not higher than 550°C; cold-

25   forming the cooled steel sheet into a tubular shape; then
applying submerged arc welding to the abutted portion
from the outer and inner sides thereof; and thereafter
subjecting the steel pipe to pipe expansion.

(16) A method for producing a high-strength steel

30   pipe excellent in low temperature toughness and weld
heat-affected zone toughness according to item (15),
characterized by heating the seam-welded portion of the
steel pipe to 300°C to 500°C before pipe expansion.

(17) A method for producing a high-strength steel

35   pipe excellent in low temperature toughness and weld
heat-affected zone toughness according to item (15),

characterized by heating the seam-welded portion of the steel pipe to 300°C to 500°C after pipe expansion.

BRIEF DESCRIPTION OF THE DRAWINGS

Figure 1 is a graph showing the influence of Nb amounts on toughness at reheated coarse grain portions.

DESCRIPTION OF THE PREFERRED EMBODIMENTS

Firstly, weld heat-affected zone toughness is explained hereunder. Two-pass welding was applied to various kinds of ultra-high-strength steels and then the toughness at welds and weld heat-affected zones at -20°C was evaluated by applying a Charpy impact test to specimens each of which had a notch at an intersection of outer and inner welds or at a portion 1 mm away from an intersection of the outer and inner welds. A mating portion means a point where the beads of double-layer weld intersect with each other on the cross section perpendicular to a welding direction. As a result of the evaluation, almost one hundred percent of all fractured surfaces were brittle fractured faces and, in some cases, the Charpy absorbed energy was low at not higher than 50 J.

As a result of investigating the fractured surfaces precisely, it was made clear that the brittle fracture originated from the following portions: (1) the region from a mating portion to a portion 1 mm away therefrom in a weld heat-affected zone that was heated once to a temperature immediately below the melting point and then reheated to a temperature immediately above the Ac_3 point, (2) the region that was reheated to a temperature immediately below the melting point, and (3) the region that was heated once to a temperature immediately below the melting point. The probability of the occurrence of brittle fracture at the respective regions was about 60% in (1), about 30% in (2), and about 10% in (3).

The result means that toughness at a reheated portion, where grains are coarsened by the influence of the one time heating, must be improved. Then, the

present inventors, as a result of observing the fractured surfaces further precisely, confirmed that Nb combined carbonitride existed at the initiation point of the brittle fracture and found the possibility of improving

5   toughness at a weld heat-affected zone, particularly at a reheated coarse grain portion that was influenced twice heat affects, by decreasing an Nb amount.

On the basis of the above findings, the present inventors investigated the influence of Nb on weld heat-

10   affected zone toughness by simulating the influence of heat caused by double layer welding through weld reproducing heat cycle test. Steel plates were produced by controlling the addition amounts of the elements other than Nb in the range specified in claim 1 or 2 and

15   varying the Nb amount in the range from 0.001 to 0.04 in terms of mass percent and test pieces were prepared. The heat cycle conditions corresponding to 2.5 kJ/mm in terms of heat input were adopted. That is, the first heat treatment was applied to the test pieces under the

20   conditions that a test piece was heated at a heating rate of 100°C/sec. to a temperature of 1,400°C, retained at the temperature for one second, and thereafter cooled at a cooling rate of 15°C/sec. in the temperature range from 500°C to 800°C, and, in addition to that, the second heat

25   treatment was applied thereto under the conditions that the heating temperature was set at 1,400°C or 900°C with the conditions of heating rate, retention time, cooling temperature and cooling rate being identical to the first heat treatment. Further, test pieces of a standard

30   dimension for V-notch Charpy impact tests were prepared in conformity with JIS Z 2202 and the Charpy impact tests were performed at -40°C in conformity with JIS Z 2242.

The results are shown in Figure 1. It was clarified that, in the steels to which Nb was added to not less

35   than 0.01%, the Charpy absorbed energy was sometimes not higher than 50 J, but, in the steels to which Nb was added to less than 0.01%, Charpy absorbed energy of not

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higher than 50 J disappeared and the toughness at a
reheated coarse grain portion remarkably improved. When
the fractured surface of a test piece of a Nb added steel
whose Charpy absorbed energy was not higher than 50 J was
5   observed, almost the entire surface was a brittle
fractured face and Nb combined carbonitride existed at
the initiation point of the brittle fractured face. On
the other hand, when the fractured surface of a steel
having an Nb content of less than 0.01% after subjected
10   to Charpy impact test was observed, no Nb combined
carbonitride existed at the initiation point of the
brittlely fractured face. Consequently, the present
inventors succeeded in improving toughness at the above-
mentioned brittle regions by reducing an Nb amount to
15   less than 0.01%.

     Next, the low temperature toughness of a base steel
is explained hereunder. It is necessary to make a
structure mainly composed of bainite and martensite
transformed from particulate unrecrystallized austenite
20   for securing an excellent low temperature toughness in a
ultra-high-strength steel pipe having a tensile strength
of not lower than 800 MPa, particularly not lower than
900 MPa. When coarse grains are mixed or the fraction of
bainite and martensite is not sufficiently high, low
25   Charpy absorbed energy is obtained, the Charpy absorbed
energy representing the property of stopping a high-speed
ductile fracture. The present inventors applied Charpy
impact tests to base steels at -60°C and precisely
investigated the structures in the vicinity of fractured
30   portions of the test pieces that could not achieve the
Charpy absorbed energy of not lower than 200 J. As a
result of the investigation, it was found that coarse
grains 10 to 100 μm in diameter existed in a structure
and they caused the reduction of Charpy absorbed energy.

35   The cast structure of a continuously cast casting
containing relatively small amount of alloying elements
and having a tensile strength of not higher than 800 MPa

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is generally composed of a composite structure of ferrite
and bainite or of ferrite and pearlite. When the casting
is reheated for hot rolling, new austenite is generated
abundantly mainly from ferrite grain boundaries and, when
5   the heating temperature is around 950°C, that is,
immediately above the Ac_3 point, the composite structure
transforms into grain adjusted austenite about 20 μm in
average grain diameter. When a steel plate is produced
through succeeding hot rolling, the structure has a finer
10   grain due to recrystallization and becomes an almost
uniform grain adjusted structure having austenite grains
about 5 μm in average diameter. However, it is estimated
that, when a steel to which elements to enhance
hardenability are added for further strengthening, like a
15   high-strength steel having a tensile strength of not
lower than 800 MPa, is hot rolled, coarse grains
partially remain and low temperature toughness
deteriorates.

    In view of this situation, the present inventors
20   investigated the influence of components on a structure
in detail and found that, when an Nb amount was reduced
to less than 0.01%, grains after hot rolling became fine
and coarse grains partially existing disappeared. The
effect of the reduction of an Nb amount can be explained
25   as follows.

    To begin with, the cause of the fact that coarse
grains partially remain when an Nb amount is large is
explained. A ultra-high-strength steel having a tensile
strength of not lower than 800 MPa, particularly not
30   lower than 900 MPa, generally contains relatively
abundantly alloying elements, such as Mn, Ni, Cu, Cr and
Mo, that provide a high hardenability. When such a steel
is produced through continuous casting or the like, the
structure of a casting after it is cooled to the room
35   temperature is made to consist of a single phase of
coarse bainite (hereunder referred to as "bainite"), the

crystal grain diameter of which is not smaller than 1 mm
in terms of prior austenite grain diameter, a single
phase of martensite (hereunder referred to as
"martensite"), or a structure mainly composed of bainite
5   and martensite (hereunder referred to as "bainite and
martensite dominant structure"). Such a structure
contains fine retained austenite in its grains. Note
that, though the structures of both bainite and
martensite are lath structures and they can hardly be
10   identified with an optical microscope, they can be
identified by hardness measurement.

When a casting having such a cast structure as
described above is heated to a temperature in the range
from 900°C to 1,000°C, the reaction of generating new
15   austenite grains by the transformation from prior
austenite grain boundaries (hereunder referred to as
"normal ferrite/austenite transformation") and the
reaction of generating coarse austenite grains not
smaller than 1 mm in size by the easy growth and
20   consolidation of the aforementioned retained austenite
(hereunder referred to as "abnormal ferrite/austenite
transformation") are generated.

When Nb is further added to such a steel, fine Nb
carbide forms and therefore the growth of grains during
25   heating is suppressed. Therefore, when a steel is heated
in the temperature range from a temperature immediately
above the Ac_3 point to 1,100°C for example, the growth of
austenite grains generated by ordinary austenite
transformation, namely secondary recrystallization, is
30   suppressed. As a result, austenite grains not smaller
than 1 mm in size, almost the same size as prior
austenite grains in a casting, are generated partially by
abnormal ferrite/austenite transformation. If such
coarse austenite grains are generated in a steel during
35   heating, as recrystallization after hot rolling hardly
occurs, the austenite grains remain partially as grains
not smaller than 50 μm in size and those coarse grains

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cause the deterioration of low temperature toughness.

When a steel is heated in the temperature range of not lower than 1,150°C, Nb combined carbide that acts as pinning grains dissolves and the growth of grains

5   generated by ordinary austenite transformation from prior austenite grain boundaries, namely secondary recrystallization, is accelerated, and, by so doing, the size of austenite grains is properly adjusted. When a casting having such a structure is hot rolled, though the

10   average grain diameter increases to some extent, coarse grains about 50 μm in size are not observed at all.

However, coarse grains smaller than about 20 μm in size still remain.

In contrast with the above, since a casting of a

15   steel wherein an Nb amount is reduced to less than 0.01% has little Nb carbide, the effect of suppressing secondary recrystallization is weak. Therefore, when the casting is heated in the temperature range from 950°C to 1,100°C, secondary recrystallization is accelerated and,

20   by so doing, the grains generated by normal austenite transformation erode coarse grains generated by abnormal ferrite/austenite transformation and the structure becomes uniform. When a casting having such a structure is hot rolled, a uniform structure having grains about 10

25   μm in average diameter is obtained and coarse grains of not smaller than 20 μm do not remain any more. Note that, as the coarsening of austenite grains after secondary recrystallization is suppressed as the heating temperature lowers, grains after hot rolling become fine.

30   As explained above, the present inventors found that, even in a casting to which alloying elements with a high hardenability were added relatively abundantly for high-strengthening and which had a single phase of bainite, a single phase of martensite, or a bainite and

35   martensite dominant structure, those being apt to generate coarse austenite grains partially by abnormal

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ferrite/austenite transformation during heating, it was
possible to conspicuously suppress the generation of
coarse grains by reducing an Nb amount to less than
0.01%. On the basis of the finding, the present
5   inventors succeeded in the development of a high-strength
steel as a base steel having excellent low temperature
toughness of not lower than 200 J in terms of Charpy
absorbed energy when the base steel was subjected to a
Charpy impact test in the temperature range from -60°C to
10   lower than -40°C.

However, it is thought that, when an Nb amount is
reduced, the recrystallization temperature lowers and
unrecrystallization rolling is not sufficiently
performed. The present inventors investigated the
15   behavior of austenite recrystallization in a steel to
which 0.005% Nb was added and a steel to which 0.012% Nb
was added, both the steels containing, in mass, 0.05% C,
0.25% Si, 2% Mn, 0.01% P, 0.001% S, 0.5% Ni, 0.1% Mo,
0.015% Ti, 0.0010% B, 0.015% Al, 0.0025% N, 0.5% Cu and
20   0.5% Cr. As a result of the investigation, it was
clarified that the recrystallization temperature of
either of the steels was in the temperature range from
900°C to 950°C regardless of the addition amount of Nb,
and, in a steel to which Mn, Ni, Cu, Cr and Mo were added
25   abundantly, the recrystallization temperature did not
change regardless of the addition of Nb. Therefore, it
was proved that it was not essential to add Nb from the
viewpoint of the recrystallization of austenite.

Further, as the reduction of an Nb amount causes the
30   decrease of strength, the present inventors studied the
addition amount of elements to enhance hardenability and
contrived to secure both strength and low temperature
toughness simultaneously by controlling a P value, that
was an index of hardenability, in an appropriate range.
35   As a result of investigating, in detail, the influence of
alloying elements on hardenability of a steel wherein an
addition amount of Nb was reduced to less than 0.01%, it

was clarified that, in the case of a steel not containing
B, by defining the P value to P = 2.7C + 0.4Si + Mn +
0.8Cr + 0.45(Ni + Cu) + 2V + Mo - 0.5, hardenability was
evaluated properly and the appropriate range of the P
5   value was from 1.9 to 3.5. On the other hand, it was
clarified that, in the case of a steel to which B is
added, the P value was defined by P = 2.7C + 0.4Si + Mn +
0.8Cr + 0.45(Ni + Cu) + 2V + 1.5Mo and the appropriate
range of the P value was from 2.5 to 4.0. By controlling
10   a P value to an appropriate range, the present inventors
succeeded in obtaining a good balance between the target
strength and low temperature toughness without the
impairment of weld heat-affected zone toughness and
weldability on site.

15   Further, when a weld heat-affected zone was heated
to a temperature of not lower than 300°C, fine
martensite-austenite (MA) was tempered and, therefore, a
high Charpy absorbed energy was obtained stably. On the
other hand, when a weld heat-affected zone of a steel to
20   which Nb was added at not less than 0.01% was heated to a
temperature of not lower than 300°C, though fine
martensite-austenite (MA) was tempered, the brittlement
occurred, at the same time, caused by the precipitation
of Nb and therefore a conspicuous effect, as expected in
25   the present invention, was not seen.

        Next, the reasons for limiting the components of a
steel plate and those of the base steel of a steel pipe
are explained hereunder.

        C is extremely effective for improving the strength
30   and hardenability of a steel by the dissolution of C or
the precipitation of carbonitride in the steel, and the
lower limit of a C content is set at 0.02% in order to
achieve a target strength by making a structure consist
of bainite, martensite, or a bainite and martensite
35   dominant structure. On the other hand, when a C content
is excessive, low temperature toughness of a steel
material and at a weld heat-affected zone deteriorates

and, thus, the weldability at a site deteriorates
conspicuously, for example low temperature cracks occur
after welding, and therefore the upper limit of a C
content is set at 0.10%. It is preferable to set the
5   upper limit of a C content at 0.07% to further improve
low temperature toughness. Here, it is preferable to
control a C content to not less than 0.03% for improving
strength. On the other hand, if strength is too high,
the shape of a steel pipe may be impaired after pipe
10   expansion and the roundness may deteriorate, and
therefore it is preferable to control the C content to
less than 0.05%. Here, roundness is obtained by
measuring the diameter of a steel pipe at plural
portions, for example measuring the diameter passing
15   through the center of a steel pipe at four portions apart
from the seam-weld at an every angle of 45 degrees,
calculating the average value, deducting the minimum
diameter from the maximum diameter, and then dividing the
deduction by the average value.
20   Si has the function of deoxidation and the effect of
enhancing strength. However, when Si is added
excessively, weld heat-affected zone toughness and
weldability on site are remarkably deteriorated and
therefore the upper limit of an Si content is set at
25   0.8%. A preferable upper limit of an Si amount is 0.6%.
Here, as Al and Ti also have the function of deoxidation,
like Si, in a steel according to the present invention,
it is preferable to adjust an Si content according to the
contents of Al and Ti. The lower limit of an Si content
30   is not particularly specified but Si is generally
contained by not less than about 0.01% as an impurity in
a steel.
        Mn is an indispensable element for making the
microstructure of a steel according to the present
35   invention consist of a bainite and martensite dominant
structure and securing a good balance between strength
and low temperature toughness, and thus the lower limit

of an Mn content is set at 1.5%. On the other hand, if
Mn is added excessively, not only hardenability is
increased and weld heat-affected zone toughness and
weldability at a site are deteriorated, but also center
5   segregation is accelerated and the low temperature
toughness of a steel material is deteriorated. For those
reasons, the upper limit of an Mn content is set at 2.5%.
Here, center segregation means the state wherein the
segregation of components generated caused by
10   solidification in the vicinity of the center of a casting
in a casting process does not disappear even after being
subjected to the subsequent processes and remains in the
vicinity of the center of the thickness of the steel
plate.
15   P and S are inevitably included impurity elements.
P accelerates center segregation and, at the same time,
improves low temperature toughness by intergranular
fracture. S lowers ductility and toughness by the
influence of MnS, that elongates during hot rolling, in a
20   steel. Therefore, in the present invention, the upper
limits of a P content and an S content are set at 0.015%
and 0.003% respectively for further improving low
temperature toughness and weld heat-affected zone
toughness. Note that, P and S are impurities and the
25   lower limits of their contents are about 0.003% and
0.0001% respectively under current technology. Further,
it is possible to suppress the precipitation of sulfide
such as MnS in a steel by restricting an S content to not
more than 0.001%. For that reason, it is preferable to
30   restrict an S content to not more than 0.001% for
suppressing the deterioration of ductility and toughness.
Ni, compared with Mn, Cr or Mo, is able to reduce a
formation of a hardened structure which is harmful to low
temperature toughness at a center segregation zone formed
35   at hot rolling. Ni is also effective to increase
toughness at weld heat-affected zone. Since the effects
are insufficient with an Ni content of less than 0.01%,

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the lower limit thereof is set at 0.01%. Further, it is
preferable to set the lower limit of an Ni content at
0.3% for the improvement of weld heat-affected zone
toughness. On the other hand, if an Ni content is
5   excessive, not only the economical efficiency
deteriorates because Ni is expensive but also weld heat-
affected zone toughness and weldability at a site
deteriorate, and therefore the upper limit of an Ni
content is set at 2.0%. Note that, the addition of Ni is
10   also effective in the prevention of surface cracks caused
by Cu during continuous casting and hot rolling. When Ni
is added for that purpose, it is preferable to add Ni to
not less than one-third of the Cu content.
          Mo is added for improving the hardenability of a
15   steel and obtaining bainite, martensite, or a bainite and
martensite dominant structure, those being excellent in a
balance between strength and low temperature toughness.
The effects are enhanced further by adding Mo in
combination with the addition of B. Further, by the
20   coexistence of Mo with B, the effects of suppressing the
recrystallization of austenite during controlled rolling
and thus fining an austenite structure are obtained. For
obtaining those effects of Mo addition, the lower limit
of an Mo content is set at 0.2% in the case of a steel to
25   which B is not added, and the same is set at 0.1% in the
case of a steel to which B is added. On the other hand,
if Mo is added in excess of 0.8%, not only a production
cost increases but also weld heat-affected zone toughness
and weldability at a site deteriorate regardless of the
30   addition of B. Therefore, the upper limit of an Mo
content is set at 0.8%. Here, a preferable upper limit
of an Mo content is 0.6%.
          Nb suppresses the recrystallization of austenite
during controlled rolling, makes an austenite structure
35   fine by the precipitation of carbonitride, and also
contribuies to the improvement of hardenability. In
particular, the effect of the improvement of

hardenability by the addition of Nb is synergistically
enhanced by its coexistence with B. However, if Nb is
added to not less than 0.01%, coarse grains are partially
generated, thus a percent fracture in an impact test is

5 lowered and weld heat-affected zone toughness is
deteriorated when double or more layer welding is
applied. Further, in that case, weldability at a site is
also deteriorated. For those reasons, the upper limit of
an Nb content is set at less than 0.01%. A preferable Nb

10 content is not more than 0.005%. Further, it is not
necessary to add Nb as long as a P value defined by the
expression $P = 2.7C + 0.4Si + Mn + 0.8Cr + 0.45(Ni + Cu)$
$+ 2V + Mo - 0.5$ is in the range from 1.9 to 4.0,
preferably from 1.9 to 3.5, in a steel not containing B

15 or a P value defined by the expression $P = 2.7C + 0.4Si +$
$Mn + 0.8Cr + 0.45(Ni + Cu) + 2V + 1.5Mo$ is in the range
from 2.5 to 4.0. However, Nb is usually contained at not
less than 0.001% in a steel as an impurity.

Ti forms fine nitride in a steel and suppresses the
20 coarsening of austenite during reheating. Further, in a
B added steel, Ti reduces dissolved N that is harmful to
the improvement of hardenability by fixing N as nitride
and thus improves hardenability further. Furthermore,
when an Al content is not more than 0.005%, Ti forms an

25 oxide in a steel. The Ti oxide functions as
intragranular transformation product nuclei at a weld
heat-affected zone and thus makes the structure of the
weld heat-affected zone fine. It is preferable to set
the lower limit of a Ti content to 0.001% for securing

30 the aforementioned effects of Ti addition. Further, it
is preferable to regulate the lower limit of a Ti content
to not less than 3.4N for stably obtaining the effects
caused by the formation of nitride and the fixation of
dissolved N. On the other hand, if an addition amount of

35 Ti is excessive, nitride coarsens, fine carbide is
generated, precipitation hardening occurs and, therefore,
weld heat-affected zone toughness is deteriorated.

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Further, in that case, as in the case where Nb is added
to not less than 0.01%, coarse grains are partially
generated and thus low temperature toughness is
deteriorated. For those reasons, the upper limit of a Ti
5   content is set at 0.030%.

Al is added in a steel as a deoxidizer and also has
the function of fining a structure. However, if the Al
content exceeds 0.1%, nonmetallic inclusions of an
aluminium oxide system increase, thus the cleanliness of
10   a steel is impaired, and also the toughness of a steel
material and at a weld heat-affected zone is
deteriorated. For those reasons, the upper limit of an
Al content is set at 0.1%. A preferable upper limit
thereof is 0.07% and the optimum Al content is not more
15   than 0.06%. Further, as Si and Ti also have the same
function of deoxidation as Al has, in a steel according
to the present invention, it is preferable to control an
Al content in consideration of the contents of Si and Ti.
The lower limit of an Al content is not specified, but Al
20   is usually contained at not less than 0.005%.

N, when it is added in excess of 0.008%, generates
surface defects on a casting and causes the deterioration
of weld heat-affected zone toughness by dissolved N and
Nb nitride. Therefore, the upper limit of an N content
25   is set at 0.008%. A preferable upper limit of an N
content is 0.006%. The lower limit of an N content is
not specified because the lower the N content, the
better, but N is usually contained at about 0.003% as an
impurity.

30   A steel according to the present invention contains
the components explained above as basic components. In
addition, for contriving to further improve strength and
toughness and expand the producible size of steel
materials, one or more of B, V, Cu, Cr, Ca, REM and Mg
35   may be added to the contents specified below.

B is an element effective in enhancing the
hardenability of a steel by adding a trace amount of B

and in obtaining a bainite and/or martensite dominant structure that is one of the objects of the present invention. Further, B enhances the effect of Mo in improving the hardenability of a steel according to the

5 present invention and accelerates the effect of improving hardenability synergistically by the coexistence of B with Nb. Those effects are not secured when the B content is less than 0.0003%. Therefore, the lower limit of a B content is set at 0.0003%. On the other hand, if

10 B is added excessively, not only is the formation of brittle grains such as $Fe_{23}(C,B)_6$ accelerated and, thus, low temperature toughness is deteriorated but, also, the effect of B in improving hardenability is impaired. Therefore, the upper limit of a B content is set at

15 0.0030%.

V has almost the same function as Nb has. Though the effects of V are weaker than those of Nb with a single addition of V, the coexistence of V with Nb further enhances the effects of improving low temperature

20 toughness and weld heat-affected zone toughness. Since those effects are insufficient with a V content of less than 0.001%, it is preferable to set the lower limit thereof to 0.001%. On the other hand, if the addition amount of V exceeds 0.3%, weld heat-affected zone

25 toughness, particularly weld heat-affected zone toughness when double or more layer welding is applied, is deteriorated, coarse grains caused by abnormal ferrite/austenite transformation during heating for hot rolling are generated, thus low temperature toughness is

30 deteriorated and, further, weldability on site is impaired. For those reasons, it is preferable to set the upper limit of a V content at 0.3%. A still preferable upper limit of a V content is 0.1%.

Cu and Cr are elements that enhance strength of a

35 base steel and at a weld heat-affected zone, and it is necessary to contain them at not less than 0.01% respectively to obtain those effects. On the other hand,

if the content of Cu or Cr is excessive, weld heat-
affected zone toughness and weldability on site are
deteriorated considerably. Therefore, each of the upper
limits of the contents of Cu and Cr is set at 1.0%.

5   Ca and REM have the functions of controlling the
shape of sulfide such as MnS in a steel and improving the
low temperature toughness of the steel. It is preferable
to set each of the lower limits of the contents of Ca and
REM at 0.0001%. On the other hand, if Ca is added in
10   excess of 0.01% or REM in excess of 0.02%, CaO-CaS or
REM-CaS is generated in large quantities, which forms
large clusters and large inclusions and, thus, the
cleanliness of a steel is impaired and weldability on
site is deteriorated. For those reasons, it is
15   preferable to set the upper limits of the contents of Ca
and REM at 0.01% and 0.02% respectively. Further, a
still preferable upper limit of a Ca content is 0.006%.

       In addition, when a strength of not lower than 950
MPa is required, it is preferable to further regulate the
20   contents of S and O in a steel to 0.001% and 0.002%
respectively. Furthermore, it is preferable to control
an ESSP value, that is an index related to the shape
control of sulfide system mixtures, (ESSP being defined
by the expression ESSP = $(Ca)[1 - 124(O)]/ 1.25S$) in the
25   range from 0.5 to 10.0.

       Mg has the functions of forming finely dispersed
oxide, suppressing the coarsening of austenite grains at
a weld heat-affected zone, and thus improving low
temperature toughness. The lower limit of an Mg content
30   is set at 0.0001% for securing those effects. On the
other hand, if an Mg content exceeds 0.006%, coarse oxide
is generated and thus low temperature toughness is
deteriorated. Therefore, the upper limit of an Mg
content is set at 0.006%.

35   In addition to the limitation on the content of each
of the addition elements, the present invention regulates
a P value, that is an index of hardenability within an

appropriate range, to obtaining an excellent balance between strength and low temperature toughness. A P value is defined differently according to the presence of B in a steel: in a steel not containing B, a P value is

5 defined by the expression $P = 2.7C + 0.4Si + Mn + 0.8Cr + 0.45(Ni + Cu) + 2V + Mo - 0.5$; and in a steel containing B, a P value is defined by the expression $P = 2.7C + 0.4Si + Mn + 0.8Cr + 0.45(Ni + Cu) + 2V + 1.5Mo$. When a P value is less than 1.9 in a steel without B addition or

10 less than 2.5 in a steel with B addition, tensile strength of not lower than 800 MPa is not obtained, and therefore those values are determined to be the lower limits in respective steels. On the other hand, when a P value exceeds 4.0 in either of the steels, weld heat-

15 affected zone toughness and weldability at a site are deteriorated, and therefore the value is determined to be the upper limit in either of the steels. Furthermore, it is preferable to determine the upper limit of a P value to be 3.5 in a steel without B addition. In conclusion,

20 an adequate range of a P value is determined to be: from 1.9 to 4.0, preferably from 1.9 to 3.5, in a steel without B addition; and from 2.5 to 4.0 in a steel with B addition.

Next, a microstructure is explained hereunder.

25 To attain a high strength of not lower than 800 MPa in terms of tensile strength and securing good low temperature toughness, it is necessary to control the amount of bainite, martensite, or a bainite and martensite dominant structure in the range from 90 to

30 100% in terms of a bainite and martensite fraction. Note that, the balance seems to be retained austenite, but it is hard to identify with an optical microscope. Here, that a bainite and martensite fraction is in the range from 90 to 100% is defined by the following two

35 conditions. Firstly, (1) confirming that polygonal ferrite is not generated by an optical micrograph, a scanning electron micrograph, or a transmission electron

micrograph, and secondly, (2) defining that a bainite and martensite fraction is in the range from 90 to 100% as follows according to hardness: to calculate the hardness of 100% martensite from the amount of C using the

5   expression $Hv = 270 + 1,300C$, wherein C is the amount of C expressed in terms of mass percent; and when the hardness of a steel material is in the range from 70 to 100% of the hardness of the 100% martensite, it is defined that the bainite and martensite fraction of the

10   steel material is in the range from 90 to 100%.

In addition, when a bainite and martensite fraction is in the range from 90 to 100%, tensile strength and a C amount satisfy the following expression: $0.7 \times (3,720C + 869) < TS$, wherein TS is tensile strength [in terms of

15   MPa] of a steel obtained and C is a C amount [in terms of mass percent].

For obtaining excellent low temperature toughness in the direction of a cross section in the case of a steel pipe for a line pipe for example, it is necessary to

20   optimize an austenite phase before the austenite phase transforms into a ferrite phase, or what is called the structure of prior austenite, at the time of cooling, and to make the final structure of a steel material efficiently fine. For that reason, prior austenite is

25   required to consist of unrecrystallized austenite and also the average grain diameter thereof is limited to not larger than 10 μm. By so doing, extremely good balance between strength and low temperature toughness is obtained. Here, the diameter of prior austenite grains

30   means the diameter of grains including a deformation band and a twin boundary that have the same function as an austenite grain boundary. The diameter of prior austenite grains is determined, for example in conformity with JIS G 0551, by dividing the full length of a

35   straight line drawn in the direction of the steel sheet thickness by the number of the points where the straight line intersects with the grain boundaries of the prior

austenite existing on the straight line, by using an
optical micrograph. The lower limit of the average
diameter of prior austenite grains is not specified, but
the detectable lower limit is about 1 μm according to a

5   test with an optical micrograph. Here, a preferable
range of a prior austenite grain diameter is from 3 to 5
μm.

In the production of a high-strength steel excellent
in low temperature toughness according to the present

10   invention, it is desirable to carry out hot rolling under
the conditions described below. A reheating temperature
is determined to be in a temperature range wherein the
structure of a casting substantially consists of a single
austenite phase, namely the Ac_3 point is determined to be

15   the lower limit of a reheating temperature. When a
reheating temperature exceeds 1,300°C, crystal grains
coarsen and, therefore, it is preferable to limit a
reheating temperature to not higher than 1,300°C. With
regard to rolling after the reheating, it is preferable

20   to firstly carry out recrystallization rolling and
secondly carry out unrecrystallization rolling. Note
that, though a recrystallization temperature varies
according to steel components, it·is in the range from
900°C to the reheating temperature, and therefore the

25   preferable temperature range during recrystallization
rolling is from 900°C to 1,100°C and the preferable
temperature range during unrecrystallization rolling is
from 750°C to 880°C. Thereafter, cooling is applied at a
cooling rate of not lower than 1°C/sec. up to an

30   arbitrary temperature of not higher than 550°C. The
upper limit of a cooling rate is not particularly
specified, but a preferable range thereof is from 10 to
40°C/sec. The lower limit of a cooling end temperature
is neither particularly specified, but a preferable range

35   thereof is from 200°C to 450°C.

By carrying out hot rolling under such conditions of

steel components, heating and rolling as explained above, a ultra-high-strength steel sheet excellent in low temperature toughness can be obtained. Further, by cold-forming the hot-rolled steel plate into a pipe and

5 thereafter applying double or more layer seam welding to an abutted portion, a ultra-high-strength steel pipe excellent in low temperature toughness and weld heat-affected zone toughness can be produced. That is, by the present invention, it is made possible to mitigate

10 welding conditions in the production of a steel pipe having such a sheet thickness when double or more layer welding is required. It is preferable to employ arc welding, particularly submerged arc welding, for seam welding.

15 The size of a high-strength steel pipe used for a line pipe according to the present invention is usually about 450 to 1,500 mm in diameter and about 10 to 40 mm in wall thickness. As a method for producing a steel pipe of such a size efficiently, the production method

20 preferably includes the processes of: producing a pipe in a UO process where a steel plate is formed into a U-shape and then into an O-shape; tack-welding the abutted portion; thereafter applying submerged arc welding from the inner and outer sides, and thereafter securing

25 roundness by pipe expansion.

Submerged arc welding is the one wherein the dilution of a weld metal by a base steel is large. Therefore, for controlling the chemical components of a weld metal in a range wherein desired properties are

30 obtained, it is necessary to select a weld material in consideration of the dilution by a base steel. As an example, welding may be carried out by using: a weld wire containing Fe as the main component, 0.01 to 0.12% C, not more than 0.3% Si, 1.2 to 2.4% Mn, 4.0 to 8.5% Ni, and

35 3.0 to 5.0 % Cr + Mo + V; and a flux of a agglomerated type or a fused type.

The ratio of dilution by a base steel varies

depending on welding conditions, particularly a weld heat input, and, in general, the ratio of dilution by a base steel increases with the increase of a heat input. However, under the condition of slow welding speed, the
5   ratio of dilution by a base steel does not increase even when a heat input increases. For securing sufficient weld penetration when one pass welding is applied to an abutted portion from the outer side and the inner side thereof, it is preferable to limit a heat input and a
10   welding speed to the following ranges.

When a heat input is less than 2.5 kJ/mm, weld penetration decreases but, on the other hand, when a heat input is larger than 5.0 kJ/mm, a weld heat-affected zone softens and weld heat-affected zone toughness somewhat
15   deteriorates. Therefore, it is preferable to limit a heat input in the range from 2.5 to 5.0 kJ/mm.

When a welding speed is lower than 1 m/min., the welding work is somewhat inefficient as seam welding for a line pipe but, on the other hand, when a welding speed
20   exceeds 3 m/min., a bead shape is hardly stable. Therefore, it is preferable to limit a welding speed in the range from 1 to 3 m/min.

Roundness can be improved by applying pipe expansion after seam welding. It is preferable to set a pipe
25   expansion rate at not less than 0.7% for improving roundness by applying plastic deformation. On the other hand, if a pipe expansion rate exceeds 2%, the toughness of both a base steel and a weld deteriorates to some extent caused by plastic deformation. For those reasons,
30   it is preferable to determine a pipe expansion rate to be in the range from 0.7 to 2%. Here, a pipe expansion rate is defined by the value obtained by subtracting a circumference before pipe expansion from a circumference after pipe expansion, dividing the resulting value by the
35   circumference before pipe expansion, and expressing the resulting value as a percentage.

After seam welding, when a seam weld is heated to

not lower than 300°C before and/or after pipe expansion,
a massive mixture of martensite and austenite (referred
to as "MA") generated at a weld heat-affected zone can be
decomposed into a bainite and martensite dominant

5   structure and fine hard cementite and, therefore, weld
heat-affected zone toughness improves. On the other
hand, if a heating temperature exceeds 500°C, a base
steel softens. For those reasons, it is preferable to
limit the heating temperature in the range from 300°C to

10   500°C. Though the influence of time is not large, it is
preferable that the time is about 30 seconds to 60
minutes. A preferable range thereof is about 30 seconds
to 50 minutes. Further, when heating is applied after
pipe expansion, a processing strain converging at the toe

15   of a weld recovers and thus weld heat-affected zone
toughness improves.

When a test piece is cut out from a weld heat-
affected zone, specularly polished and etched, and then
observed with a scanning electron microscope, it is seen

20   that an MA formed at a weld heat-affected zone is
entirely composed of a white massive substance. When an
MA is heated to 300°C to 500°C, it is decomposed into a
bainite and martensite dominant structure having fine
precipitates in the grains and cementite, and these can

25   be distinguished from the MA. Further, when a test piece
is subjected to repeller etching or nitral etching after
specularly polished and observed with a scanning electron
microscope too, an MA can be distinguished from another
MA decomposed into a bainite and martensite dominant

30   structure and cementite by judging whether or not fine
precipitates exist in grains.

Here, when a seam weld is heated, it is preferable
to apply the heating to a weld metal and the weld heat-
affected zone of a base steel. A weld heat-affected zone

35   is the area within about 3 mm from an intersection of a
weld metal and a base steel and therefore it is
preferable to heat at least the area including a base

steel within 3 mm from an intersection of a weld metal
and a base steel. However, it is technically difficult
to heat such a narrow area and therefore it is realistic
to apply a heat treatment to the area within about 50 mm
5 from an intersection of a weld metal and a base steel.
Here, there is no inconvenience such as the deterioration
of base steel properties caused by a heating to a
temperature in the range from 300°C to 500°C. A gas
burner of a radiation type or an induction heater can be
10 adopted for the heating of a seam weld.

As it has been explained above, the present
invention makes it possible to produce a ultra-high-
strength steel plate having a tensile strength of not
lower than 800 MPa and a steel pipe made thereof: the
15 steel plate being excellent in weld heat-affected zone
toughness when double or more layer welding is applied;
Charpy absorbed energy of the base steel in the
temperature range of not higher than -40°C being not
lower than 200 J on the average and little dispersing;
20 the steel plate having excellent low temperature
toughness; and further the steel plate being excellent in
weldability at a site. By so doing, it is made possible
to apply the steel plate and the steel pipe for a line
pipe for the transport of natural gas or crude oil, a
25 steel plate for water pumping, a pressure vessel, a
welded structure or the like, these being used in harsh
environments.

Example 1
Steels containing chemical compositions shown in
30 Tables 1 and 2 (Table 2 being continued from Table 1)
were melted and continuously cast into castings 240 mm in
thickness. The resulting castings were reheated to
1,100°C, thereafter rolled in the recrystallization
temperature range from 900°C to 1,100°C, further rolled
35 in the uncrystallization temperature range from 750°C to
880°C, and thereafter cooled at a cooling rate of 5 to
50°C/sec. up to a temperature not higher than 420°C by

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water cooling, and by so doing, steel plates 10 to 20 mm in thickness were produced.

An average diameter of prior austenite grains was obtained by the straight line crossing segment method in the thickness direction conformity with JIS G 0551. A bainite and martensite fraction was obtained by the following procedures. To begin with, it was confirmed that polygonal ferrite was not generated by observing a structure in an optical micrograph in conformity with JIS G 0551. Then, the Vickers hardness was measured imposing a weight of 1 Kg and the measured value was defined as Hv_{BM} in conformity with JIS Z 2244. The ratio α_{BM} of Hv_{BM} to the hardness of 100% martensite calculated by the expression $Hv = 270 + 1,300C$, namely $Hv_{BM}/Hv = \alpha_{BM}$, was obtained. Thereafter, using the definition of a bainite and martensite fraction being 90% in the event of $\alpha_{BM} = 0.7$ and the same being 100% in the event of $\alpha_{BM} = 1$, a bainite and martensite fraction F_{BM} was calculated by the expression $F_{BM} = 100 \times (1/3 \times \alpha_{BM} + 2/3)$.

Yield strength and tensile strength in the direction of the rolling of a steel plate (hereunder referred to as "L direction") and in the direction perpendicular to the rolling direction (hereunder referred to as "C direction") were evaluated by the API full thickness tensile test. A Charpy impact test was carried out at -40°C with the test repetition frequency n being three in conformity with JIS Z 2242 by using V-notched test pieces of a standard size, the length of the test pieces being in the L and C directions, prepared in conformity with JIS Z 2202. A Charpy absorbed energy was evaluated as the average of the values obtained by the three repeated measurements. In addition, another Charpy impact test was carried out in the temperature range from -60°C to lower than -40°C with the test repetition frequency n varied from 3 to 30, and the probability that a Charpy

absorbed energy is not lower than 200 J (hereunder referred to as "low temperature toughness reliability") was evaluated in terms of percentage.

Weld heat-affected zone toughness was evaluated by
5   subjecting a specimen to heat treatments corresponding to welding twice, each welding having a heat input of 2.5 kJ/mm, using a weld reproducing heat cycle test apparatus. That is, the first heat treatment was applied to a specimen under the conditions that the specimen was
10   heated at a heating rate of 100°C/sec. to a temperature of 1,400°C, retained at the temperature for one second, and thereafter cooled at a cooling rate of 15°C/sec. in the temperature range from 500°C to 800°C, and, in addition to that, the second heat treatment was applied
15   thereto under the conditions that the heating temperature was set at 1,400°C or 900°C with the conditions of heating rate, retention time, cooling temperature and cooling rate being identical to the first heat treatment. Further, V-notched test pieces of standard dimension were
20   prepared in conformity with JIS Z 2202, and the Charpy impact test was applied to the test pieces at -30°C with the repetition frequency n being three in conformity with JIS Z 2242, and Charpy absorbed energy was evaluated by the average of the values obtained by the three repeated
25   measurements.

The results are shown in Table 3. Steels A to E are the ones that contain components within the ranges specified in the present invention and fulfill the target levels of strength, low temperature toughness and weld
30   heat-affected zone toughness. On the other hand, steel F has a C amount and steel I an Mn amount smaller than those in the ranges specified in the present invention and therefore the strength is low. Steel G has a C amount, steel H an Si amount, steel J an Mn amount, and
35   steel K an Mo amount larger than those in the ranges specified in the present invention and therefore low temperature toughness, low temperature toughness

reliability and weld heat-affected zone toughness are deteriorated. Steel L has an Nb amount larger than that in the range specified in the present invention, and therefore, though the Charpy absorbed energy at -40°C is good, low temperature toughness reliability and weld heat-affected zone toughness are deteriorated. Steel M has a still larger Nb amount than steel L and therefore low temperature toughness, low temperature toughness reliability and weld heat-affected zone toughness are deteriorated. Steels N, O, P and R have a Ti amount, a V amount, an N amount and an S amount, respectively, larger than those in the ranges specified in the present invention, and therefore, low temperature toughness, low temperature toughness reliability and weld heat-affected zone toughness are deteriorated. Steel Q has an Al amount larger than that in the range specified in the present invention and therefore weld heat-affected zone toughness is deteriorated.

Table 1: Chemical components (mass percent), Ceq
and Pcm of steel material

Steel	Chemical components (mass percent)										
	C	Si	Mn	P	S	Ni	Mo	Nb	Ti	Al	N
A	0.03	0.10	1.95	0.005	0.0005	0.50	0.30	0.005	0.008	0.015	0.0023
B	0.05	0.25	1.85	0.008	0.0006	0.90	0.45	0.007	0.005	0.020	0.0015
C	0.04	0.15	1.90	0.003	0.0008	2.00	0.20	0.009	0.010	0.008	0.0030
D	0.06	0.25	1.90	0.004	0.0003	1.80	0.40	0.003	0.009	0.010	0.0025
E	0.05	0.10	1.96	0.004	0.0010	1.00	0.10	0.009	0.005	0.020	0.0015
F	0.01	0.25	1.85	0.005	0.0010	1.20	0.35	0.004	0.011	0.015	0.0032
G	0.15	0.15	1.95	0.007	0.0006	0.60	0.26	0.007	0.011	0.012	0.0033
H	0.07	1.00	2.12	0.009	0.0018	0.30	0.48	0.009	0.011	0.023	0.0032
I	0.04	0.26	1.00	0.010	0.0026	0.50	0.52	0.002	0.009	0.015	0.0025
J	0.05	0.35	3.00	0.006	0.0003	0.32	0.42	0.001	0.005	0.026	0.0016
K	0.09	0.48	2.05	0.008	0.0005	0.85	1.00	0.005	0.010	0.023	0.0030
L	0.04	0.55	1.98	0.009	0.0016	0.13	0.26	0.050	0.010	0.015	0.0028
M	0.04	0.55	1.96	0.009	0.0016	0.13	0.26	0.150	0.010	0.015	0.0028
N	0.03	0.49	1.91	0.005	0.0006	0.45	0.32	0.003	0.035	0.010	0.0016
O	0.07	0.15	2.00	0.006	0.0007	0.50	0.23	0.002	0.012	0.030	0.0035
P	0.08	0.05	2.16	0.007	0.0009	0.16	0.51	0.005	0.015	0.026	0.0080
Q	0.05	0.16	1.79	0.009	0.0005	0.65	0.45	0.006	0.012	0.060	0.0035
R	0.04	0.20	1.95	0.007	0.0040	0.80	0.30	0.008	0.010	0.001	0.0030

A bar - in a cell of a chemical component means that the amount thereof is not larger than the detectable limit.
Ceq = C/ + Mn/6 + (Ni + Cu)/5 + (Mo + V + Cr)/5

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Pcm = C + Si/30 + (Mn + Cu + Cr)/20 + Ni/60 + Mo/15
    + V/10 + B + 5
P value = 2.7C + 0.4Si + Mn + 0.8Cr + 0.45(Ni + Cu)
        + 2V + Mo - 0.5(B free steel)
5   P value = 2.7C + 0.4Si + Mn + 0.8Cr + 0.45(Ni + Cu)
        + 2V + 1.5Mo(B containing steel)

Table 2 - (continued from Table 1)

Chemical components (mass percent)							P-value	Ceq	Pcm
B	V	Cu	Cr	Ca	REM	Mg			
0.0010	0.080	0.30	0.30	-	-	-	3.24	0.540	0.200
-	-	0.50	0.60	0.0012	-	0.0010	3.15	0.662	0.251
0.0023	0.040	-	-	-	0.0008	-	3.35	0.538	0.202
-	0.050	0.30	0.30	-	-	-	3.35	0.667	0.255
0.0010	-	-	0.60	-	-	-	3.22	0.583	0.210
0.0010	0.030	0.30	0.30	-	-	0.0005	3.48	0.554	0.192
0.0008	-	0.23	0.50	-	-	-	3.58	0.682	0.320
-	0.040	0.16	0.50	-	0.0008	-	3.38	0.658	0.283
0.0008	0.030	0.65	0.32	-	-	-	2.83	0.457	0.197
0.0015	0.026	0.26	0.45	0.002	-	-	4.58	0.768	0.291
0.0016	-	0.32	0.26	-	-	-	4.72	0.762	0.326
0.0013	0.050	0.15	0.52	-	-	-	3.32	0.551	0.221
0.0013	0.050	0.15	0.52	-	-	-	3.32	0.551	0.221
0.0010	0.030	0.51	0.23	0.0023	-	0.0002	3.34	0.528	0.215
-	0.150	0.30	0.42	-	-	-	2.98	0.617	0.250
0.0026	0.040	0.23	0.26	-	0.0005	-	3.82	0.628	0.268
0.0023	-	0.32	0.59	0.0021	-	-	3.57	0.621	0.243
0.0008	0.050	0.30	0.30	-	-	-	3.42	0.568	0.217

Table 3: Evaluation result

Performance No.	Steel	Mechanical properties of steel material			Low temperature toughness reliability, %	Average prior austenite grain diameter, μm	Bainite and martensite fraction, %	Weld heat-affected zone toughness	
		YS MPa	TS MPa	vE-40 J				Condition 1, J	Condition 2, J
1	A	890	1000	280	95	4.5	100	130	110
2	B	917	1030	230	93	5.1	100	125	105
3	C	872	980	290	96	4.2	100	162	142
4	D	904	1016	245	97	6.1	100	156	136
5	E	908	1020	225	94	4.6	100	166	146
6	F	703	790	210	86	7.5	81	140	130
7	G	1113	1250	56	75	4.6	100	65	45
8	H	966	1085	80	79	7.5	100	43	23
9	I	694	780	205	86	4.3	80	130	110
10	J	1077	1210	60	75	4.5	100	80	60
11	K	1028	1155	150	82	5.3	100	82	62
12	L	908	1020	205	84	6.2	100	75	55
13	M	1086	1220	80	80	4.3	100	45	25
14	N	890	1000	85	82	5.2	100	65	45
15	O	1041	1170	150	82	4.6	100	75	55
16	P	1050	1180	130	82	5.2	100	50	30
17	Q	908	1020	200	85	3.8	100	66	46
18	R	890	1000	80	80	3.2	100	45	25

vE-40 is Charpy absorbed energy measured at -40°C with the repetition frequency n being three.

Weld heat-affected zone toughness is Charpy absorbed energy measured at -30°C with the repetition frequency n being three.

The conditions of weld heat-affected zone toughness are identical to the heat treatment conditions of weld reproducing heat cycle test.

Heating rate: 100°C/sec., retention time: 1 sec., cooling temperature range: 500°C
to 800°C, cooling rate: 15°C/sec.

Condition 1: both first and second heating temperatures are 1,400°C.
Condition 2: first heating temperature is 1,400°C and second is 900°C.

Example 2

Steel plates 10 to 20 mm in thickness, the steel sheets containing the chemical components of steels A to E shown in Tables 1 and 2, were produced under the same
5   conditions as Example 1. Thereafter, the steel plates were subjected to cold forming, then submerged arc welding at a heat input of 2.0 to 3.0 kJ/mm on each of the inner surfaces and at a heat input of 2.0 to 3.0 kJ/mm on each of the outer surfaces, thereafter pipe
10   expansion, and, by so doing, steel pipes 700 to 920 mm in outer diameter were produced. An average diameter of prior austenite grains and a bainite and martensite fraction in the base steel of each of the steel pipes were obtained in the same manner as Example 1. Further,
15   tensile properties of each of the steels were evaluated by the API full thickness tensile test. Low temperature toughness was evaluated, as in Example 1, by the average value of absorbed energy and the low temperature toughness reliability of a Charpy impact test piece
20   prepared so that the length thereof may be in the C direction. Weld heat-affected zone toughness was evaluated by subjecting a test piece having a notch at an intersection or a portion 1 mm apart from an intersection to another Charpy impact test at -30°C.
25   The results are shown in Table 4. In any of the steels, the tensile strength of the base steel is not lower than 800 MPa, the toughness of the base steel is extremely good; the Charpy absorbed energy at -40°C is not lower than 200 J and the low temperature toughness
30   reliability is not less than 85%. With respect to a weld heat-affected zone, the Charpy absorbed energy at -30°C is not lower than 100 J and the weld heat-affected zone toughness is also excellent.

Table 4: Evaluation result

| Performance No. | Steel | Mechanical properties of steel pipe | | | Low temperature toughness reliability, % | Average prior austenite grain diameter, μm | Bainite and martensite fraction, % | Weld heat-affected zone toughness | |
		YS MPa	TS MPa	vE-40 J				Notch position 1, J	Notch position 2, J
19	A	917	1030	260	90	4.5	100	205	175
20	B	935	1050	210	93	5.1	100	180	150
21	C	890	1000	270	94	4.2	100	162	132
22	D	922	1036	225	90	6.1	100	145	115
23	E	935	1050	215	86	4.6	100	148	118

Notch position 1 means a mating portion and notch position 2 means a portion 1 mm away from a mating portion.

Example 3

In the same manner as Example 1, castings were
produced from a steel containing chemical components of
steel A shown in Tables 1 and 2 and, thereafter, the
5   castings were hot rolled under the conditions shown in
Table 5 and cooled and, by so doing, steel plates 10 to
20 mm in thickness were produced. In the same manner as
Example 1, an average diameter of prior austenite grains
and a bainite and martensite fraction were obtained, and
10   tensile properties were evaluated by the API full
thickness tensile test. Low temperature toughness was
evaluated, as for Example 1, by the average value of
absorbed energy and the low temperature toughness
reliability of a Charpy impact test piece prepared so
15   that the length thereof may be in the C direction. Weld
heat-affected zone toughness was evaluated by subjecting
a test piece to a weld-reproducing heat cycle test and
then a Charpy impact test at -30°C.

The results are shown in Table 6. In any of the
20   steels, the tensile strength of the base steel is not
lower than 800 MPa, with respect to the toughness of the
base steel, the Charpy absorbed energy at -40°C is not
lower than 200 J and the low temperature toughness
reliability is not less than 85% and, with respect to the
25   weld heat-affected zone, the Charpy absorbed energy at
-30°C is not lower than 100 J and, therefore, an ultra-
high-strength steel plate excellent in weld heat-affected
zone toughness is obtained. Further, steels 27 and 28
produced under the conditions in the ranges specified in
30   claim 6 have more excellent low temperature toughness
reliability than steels 24 to 26 produced under
conditions different from those specified in claim 6.

Table 5: Hot-rolling condition

Performance No.	Steel	Hot-rolling condition						
		Heating temperature, °C	Recrystallization rolling		Unrecrystallization rolling		Water cooling	
			Start temperature, °C	End temperature, °C	Start temperature, °C	End temperature, °C	Stop temperature, °C	Cooling rate, °C/sec.
24	A	1350	900	850	830	730	400	20
25	A	1100	980	930	850	810	600	20
26	A	1100	960	900	880	800	200	0.5
27	A	1150	970	930	850	790	350	40
28	A	1100	1000	950	870	780	400	15

Table 6: Evaluation result

Performance No.	Mechanical properties of steel material				Low temperature toughness reliability, %	Average prior austenite grain diameter, μm	Bainite and martensite fraction, %	Weld heat-affected zone toughness	
	Steel	YS MPa	TS MPa	vE-40 J				Condition 1, J	Condition 2, J
24	A	872	980	210	85	5.1	100	205	175
25	A	792	890	205	88	4.5	93	210	180
26	A	757	850	210	87	4.8	91	202	172
27	A	890	1000	260	95	3.5	100	223	193
28	A	872	980	250	96	4.2	100	215	185

vE-40 is Charpy absorbed energy measured at -40°C with the repetition frequency n being three.

Weld heat-affected zone toughness is Charpy absorbed energy measured at -30°C with the repetition frequency n being three.

The conditions of weld heat-affected zone toughness are identical to the heat treatment conditions of weld reproducing heat cycle test.

Heating rate: 100°C/sec., retention time: 1 sec., cooling temperature range: 500°C to 800°C, cooling rate: 15°C/sec.

Condition 1: both first and second heating temperatures are 1,400°C.
Condition 2: first heating temperature is 1,400°C and second is 900°C.

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Example 4

Steels containing chemical compositions shown in
Table 7 were melted and continuously cast into castings.
The resulting castings were reheated to 1,100°C,
5 thereafter rolled in the recrystallization temperature
range from 900°C to 1,100°C, further rolled at a
reduction ratio of 5 in the uncrystallization temperature
range from 750°C to 880°C, and thereafter cooled at a
cooling rate of 5 to 50°C/sec. up to a temperature not
10 higher than 420°C by water cooling and, by so doing,
steel sheets 16 mm in thickness were produced. An
average diameter of prior austenite grains was obtained
by the straight line crossing segment method in
conformity with JIS G 0551.

15 Yield strength and tensile strength in the C
direction of a steel sheet were evaluated by the API full
thickness tensile test. A Charpy absorbed energy was
evaluated by carrying out a Charpy impact test at -40°C
with the test repetition frequency n being three in
20 conformity with JIS Z 2242 by using V-notched test pieces
of a standard size, the length of the test pieces being
in the C direction, prepared in conformity with JIS Z
2202. Weld heat-affected zone toughness was evaluated in
the same manner as Example 1. In addition, for
25 simulating HAZ thermal cycle, specimens were subjected to
heat treatment twice, then heated to 350°C and held for
five minutes at the temperature.

Further, the value TS/0.7(3,720C + 869) was
calculated from a value of tensile strength and a C
30 amount. When a bainite and martensite fraction is within
the range from 90 to 100%, the following expression is
satisfied;

$$TS/(3,720C + 869) > 0.7,$$

wherein TS is tensile strength of a steel obtained (in
35 terms of MPa) and C is a C amount (in terms of mass
percent).

In Table 8, steels AA to AF, AH, AJ, AK, and AP to

AR are the ones that contain components within the ranges
specified in the present invention, and have the target
levels of strength, low temperature toughness and weld
heat-affected zone toughness. On the other hand, steel

5 AG has a C amount larger than that in the range specified
in the present invention and therefore the low
temperature toughness of the base steel and the weld
heat-affected zone toughness are deteriorated. Further,
steel AI has an Mn amount smaller than that in the range

10 specified in the present invention and therefore the
microstructure does not consist of a bainite and
martensite dominant structure and the strength and the
low temperature toughness are deteriorated. Steels AL
and AM have an Nb amount and steel AN a Ti amount larger

15 than those in the ranges specified in the present
invention and, therefore, coarse crystal grains are
partially generated, the Charpy absorbed energy of the
base steel is deteriorated in some of the test pieces,
and also the weld heat-affected zone toughness is

20 deteriorated. Steel AO has a P value smaller than that
in the range specified in the present invention and
therefore the tensile strength is deteriorated.

Table 7: Chemical components of steel material (mass percent)

Steel	Chemical components (mass percent)																		P-value
	C	Si	Mn	P	S	Ni	Mo	Nb	Ti	Al	N	B	V	Cu	Cr	Ca	REM	Mg	
AA	0.03	0.10	1.95	0.005	0.0005	0.50	0.30	0.005	0.009	0.015	0.0023	0.0011	0.060	0.30	0.30	-	-	-	3.24
AB	0.05	0.25	1.85	0.008	0.0006	0.90	0.45	0.007	0.005	0.020	0.0015	-	-	0.50	0.60	0.0012	-	-	3.15
AC	0.04	0.15	1.90	0.003	0.0008	1.91	0.20	0.004	0.012	0.008	0.0030	0.0017	-	-	-	-	-	0.0010	3.23
AD	0.06	0.25	1.90	0.004	0.0003	1.80	0.40	0.003	0.009	0.010	0.0025	-	0.050	0.30	0.30	-	-	-	3.35
AE	0.05	0.10	1.96	0.004	0.0010	1.00	0.10	0.009	0.007	0.020	0.0016	0.0010	-	-	0.60	-	-	-	3.22
AF	0.02	0.25	1.85	0.005	0.0010	1.20	0.35	0.004	0.011	0.015	0.0032	0.0013	0.030	0.30	0.30	-	-	-	3.50
AG	0.15	0.15	1.60	0.007	0.0006	0.60	0.26	0.007	0.014	0.012	0.0033	0.0008	-	0.00	0.00	-	-	0.0005	2.73
AH	0.07	0.54	2.12	0.009	0.0018	0.30	0.48	0.009	0.014	0.023	0.0032	-	0.040	0.16	0.50	-	-	-	3.19
AI	0.04	0.26	1.00	0.010	0.0026	0.50	0.52	0.002	0.009	0.015	0.0025	0.0008	0.030	0.65	0.32	-	0.0008	-	2.83
AJ	0.05	0.35	2.20	0.006	0.0003	0.32	0.42	0.001	0.005	0.026	0.0016	0.0015	0.026	0.26	0.45	-	-	-	3.78
AK	0.09	0.48	2.05	0.008	0.0005	0.85	0.64	0.005	0.013	0.023	0.0030	0.0000	-	0.00	0.00	0.002	-	-	3.83
AL	0.04	0.55	1.96	0.009	0.0016	0.13	0.26	0.042	0.012	0.015	0.0028	0.0013	0.050	0.15	0.52	-	-	-	3.32
AM	0.04	0.55	1.96	0.009	0.0016	0.13	0.26	0.079	0.014	0.015	0.0028	0.0013	0.050	0.15	0.52	-	-	-	3.32
AN	0.03	0.48	1.91	0.005	0.0006	0.45	0.32	0.003	0.064	0.010	0.0016	0.0010	0.030	0.51	0.23	0.0023	-	0.0002	3.34
AO	0.04	0.15	1.60	0.006	0.0007	0.10	0.23	0.002	0.012	0.030	0.0035	-	0.000	0.30	0.00	-	-	-	1.68
AP	0.08	0.05	2.20	0.007	0.0009	0.16	0.51	0.005	0.015	0.026	0.0025	0.0000	0.240	0.00	0.00	-	-	-	3.75
AQ	0.04	0.16	1.79	0.009	0.0005	0.65	0.72	0.006	0.013	0.062	0.0035	0.0016	-	0.00	0.00	0.0021	0.0005	-	3.33
AR	0.04	0.20	1.95	0.007	0.0012	0.80	0.30	0.008	0.012	0.001	0.0030	0.0008	0.000	0.00	0.00	-	-	-	2.95

A bar - in a cell of a chemical component means that the amount thereof is not larger than the detectable limit.

P value = $2.7C + 0.4Si + Mn + 0.8Cr + 0.45(Ni + Cu) + 2V + Mo - 0.5$ (B free steel)

P value = $2.7C + 0.4Si + Mn + 0.8Cr + 0.45(Ni + Cu) + 2V + 1.5Mo$ (B containing steel)

Table 8: Test conditions and evaluation result

Performance No.	Steel	Mechanical properties of steel				Average prior austenite grain diameter, μm	TS/ (3720[C] + 869)	Weld heat-affected zone toughness			
		TS	vE-40					Condition 1	Condition 2	Condition 3	Condition 4
		MPa	J	J	J			J	J	J	J
29	AA	930	280	276	281	4.5	0.948	134	110	138	162
30	AB	946	230	241	208	5.1	0.897	133	105	154	129
31	AC	980	290	277	281	4.2	0.963	156	142	180	174
32	AD	1016	221	205	198	6.1	0.930	153	136	175	151
33	AE	1020	225	204	212	4.6	0.967	165	146	179	188
34	AF	830	233	245	242	7.5	0.880	141	130	159	148
35	AG	1202	96	102	111	4.6	0.842	51	23	51	34
36	AH	1021	211	203	199	7.5	0.904	117	101	127	127
37	AI	698	164	157	123	4.3	0.686	87	110	100	111
38	AJ	1012	214	222	231	4.5	0.959	122	102	147	176
39	AK	1155	203	201	203	5.3	0.959	124	103	151	142
40	AL	987	242	64	202	6.2	0.970	87	55	88	43
41	AM	998	228	76	102	4.3	0.981	67	25	60	25
42	AN	931	85	243	221	5.2	0.949	65	45	64	22
43	AO	724	156	143	161	4.6	0.711	69	55	88	57
44	AP	977	201	210	206	5.2	0.837	128	102	147	133
45	AQ	994	248	223	254	3.8	0.977	134	134	160	179
46	AR	961	222	218	218	3.2	0.944	108	121	162	177

vE-40 is Charpy absorbed energy measured at -40°C with the repetition frequency n being three.

Weld heat-affected zone toughness is Charpy absorbed energy measured at -30°C with the repetition frequency n being three.

The conditions of weld heat-affected zone toughness are identical to the heat treatment conditions of weld reproducing heat cycle test.

Heating rate: 100°C/sec., retention time: 1 sec., cooling temperature range: 500°C to 800°C, cooling rate: 15°C/sec.

Condition 1: both first and second heating temperatures are 1,400°C.
Condition 2: first heating temperature is 1,400°C and second is 900°C.
Condition 3: both first and second heating temperatures are 1,400°C, and then heated to 350°C and retained for 5 minutes.
Condition 4: first heating temperature is 1,400°C and second is 900°C, and then heated to 350°C and retained for 5 minutes.

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Example 5

The steel plates containing the chemical components of steels AA to AE shown in Table 7 were produced in the same manner as Example 4, then formed into pipes in a UO

5   process, and subjected to submerged arc welding at a heat input of 2.0 to 3.0 kJ/mm on each of the inner surfaces and at a heat input of 2.0 to 3.0 kJ/mm on each of the outer surfaces. Subsequently, some of the steel pipes were heated to 350°C at the seam welds by induction

10   heating and then held for five minutes, and thereafter cooled to the room temperature and subjected to pipe expansion, while some of the steel pipes were subjected to pipe expansion without heating the seam welds.

For investigating the mechanical properties of the

15   base steels of those steel pipes, in the same manner as in Example 4, an API full thickness tensile test and a Charpy impact test were carried out, the Charpy impact test being carried out at -40°C using test pieces having the length in the C direction. The Charpy absorbed

20   energy was obtained by measuring it with the repetition frequency n being three and averaging the three measured values. Further, weld heat-affected zone toughness was obtained by carrying out another Charpy impact test at -30°C with the repetition frequency n being three using

25   test pieces each having a notch at an intersection or a portion 1 mm apart from an intersection and then averaging the resulting values.

The results are shown in Table 9. In Table 9, "As welded" in the column "Weld heat-affected zone toughness"

30   represents the weld heat-affected zone toughness of a steel pipe subjected to pipe expansion without the heating of a seam weld and "Heat treatment" represents the weld heat-affected zone toughness of a steel pipe subjected to pipe expansion after a seam weld is heated

35   by induction heating. In any of steels AA to AE, the tensile strength of the base steel is not lower than 900 MPa and, with respect to the toughness of the base steel,

the Charpy absorbed energy at -40°C is not lower than 200 J, and with respect to the toughness at the weld heat-affected zone, the Charpy absorbed energy at -30°C is not lower than 100 J. Therefore, high-strength steel pipes
5 excellent in the low temperature toughness of the base steel and weld heat-affected zone toughness are obtained.

Table 9: Test conditions and evaluation result

Performance No.	Steel	Mechanical properties of steel pipe			Prior austenite	TS/ (3720[C] + 869)	Weld heat-affected zone toughness			
		YS	TS	vE-40			As welded		Heat treatment	
							Notch position	Notch position	Notch position 1,	Notch position 2,
		MPa	MPa	J	μm		J	J	J	J
47	AA	867	963	278	4.5	0.948	178	165	205	184
48	AB	904	972	228	5.1	0.897	180	150	181	178
49	AC	940	1001	282	4.2	0.963	162	132	175	178
50	AD	963	1036	207	6.1	0.93	145	115	169	169
51	AE	988	1040	217	4.6	0.967	148	118	165	169

Notch position 1 means a mating portion and notch position 2 means a portion 1 mm away from a mating portion.

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CLAIMS

1.   A high-strength steel excellent in low
temperature toughness and weld heat-affected zone
toughness, characterized by containing, in mass,

5   C: 0.02 to 0.10%,
           Si: not more than 0.6%,
           Mn: 1.5 to 2.5%,
           P: not more than 0.015%,
           S: not more than 0.003%,

10   Ni: 0.01 to 2.0%,
           Mo: 0.2 to 0.6%,
           Nb: less than 0.010%,
           Ti: not more than 0.030%,
           Al: not more than 0.070%, and

15   N: not more than 0.0060%,
with the balance consisting of Fe and unavoidable
impurities; the P value of the steel defined by the
following expression being in the range from 1.9 to 3.5;
and the microstructure of the steel being mainly composed

20 of martensite and bainite:
           P = 2.7C + 0.4Si + Mn + 0.8Cr + 0.45(Ni + Cu) + 2V +
Mo - 0.5.

2.   A high-strength steel excellent in low
temperature toughness and weld heat-affected zone

25 toughness, characterized by containing, in mass,
           C: 0.02 to 0.10%,
           Si: not more than 0.6%,
           Mn: 1.5 to 2.5%,
           P: not more than 0.015%,

30   S: not more than 0.003%,
           Ni: 0.01 to 2.0%,
           Mo: 0.1 to 0.6%,
           Nb: less than 0.010%,
           Ti: not more than 0.030%,

35   B: 0.0003 to 0.0030%,
           Al: not more than 0.070%, and
           N: not more than 0.0060%, so as to satisfy the

expression Ti - 3.4N $\geqq$ 0,

with the balance consisting of Fe and unavoidable
impurities; the P value of the steel defined by the
following expression being in the range from 2.5 to 4.0;

5   and the microstructure of the steel being composed of
martensite and bainite:

P = 2.7C + 0.4Si + Mn + 0.8Cr + 0.45(Ni + Cu) + 2V +
1.5Mo.

3.   A high-strength steel excellent in low

10   temperature toughness and weld heat-affected zone
toughness according to claim 1 or 2, characterized by
further containing, in mass, one or more of

V: 0.001 to 0.10%,

Cu: 0.01 to 1.0%,

15   Cr: 0.01 to 1.0%,

Ca: 0.0001 to 0.01%,

REM: 0.0001 to 0.02%, and

Mg: 0.0001 to 0.006%.

4.   A high-strength steel excellent in low

20   temperature toughness and weld heat-affected zone
toughness according to any one of claims 1 to 3,
characterized by the average diameter of the prior
austenite grains in the steel being not larger than 10
μm.

25   5.   A high-strength steel excellent in low
temperature toughness and weld heat-affected zone
toughness, characterized by: containing, in mass,

C: 0.02 to less than 0.05%,

Si: not more than 0.6%,

30   Mn: 1.5 to 2.5%,

P: not more than 0.015%,

S: not more than 0.001%,

Ni: 0.01 to 2.0%,

Mo: 0.1 to 0.6%,

35   Nb: less than 0.010%,

Ti: not more than 0.030%,

B: 0.0003 to 0.0030%,

Al: not more than 0.070%, and

N: not more than 0.0060%, so as to satisfy the

expression Ti - 3.4N ≧ 0, and further one or more of

5   V: 0.001 to 0.10%,

Cu: 0.01 to 1.0%, and

Cr: 0.01 to 1.0%,

with the balance consisting of Fe and unavoidable

impurities; the P value of the steel defined by the

10   following expression being in the range from 2.5 to 4.0;

the microstructure of the steel being composed of

martensite and bainite; and the average diameter of the

prior austenite grains in the steel being not larger than

10 μm:

15   P = 2.7C + 0.4Si + Mn + 0.8Cr + 0.45(Ni + Cu) + 2V +

1.5Mo.

6.   A high-strength steel excellent in low

temperature toughness and weld heat-affected zone

toughness, characterized by: containing, in mass,

20   C: 0.02 to less than 0.05%,

Si: not more than 0.6%,

Mn: 1.5 to 2.5%,

P: not more than 0.015%,

S: not more than 0.003%,

25   Ni: 0.01 to 2.0%,

Mo: 0.1 to 0.6%,

Nb: less than 0.010%,

Ti: not more than 0.030%,

B: 0.0003 to 0.0030%,

30   Al: not more than 0.070%, and

N: not more than 0.0060%, so as to satisfy the

expression Ti - 3.4N ≧ 0, and further one or more of

V: 0.001 to 0.10%,

Cu: 0.01 to 1.0%,

35   Cr: 0.01 to 1.0%, and

Ca: 0.0001 to 0.01%,

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with the balance consisting of Fe and unavoidable
impurities; the P value of the steel defined by the
following expression being in the range from 2.5 to 4.0;
the microstructure of the steel being composed of
5   martensite and bainite; and the average diameter of the
prior austenite grains in the steel being not larger than
10 μm:

    P = 2.7C + 0.4Si + Mn + 0.8Cr + 0.45(Ni + Cu) + 2V +
1.5Mo.

10   7.   A method for producing a high-strength steel
sheet excellent in low temperature toughness and weld
heat-affected zone toughness, the method being the one
for producing a steel plate from a casting containing
components according to any one of claims 1 to 3, 5 and
15   6, characterized by: reheating the casting to a
temperature of not lower than the Ac_3 point; hot rolling
it; and thereafter cooling the resulting steel sheet at a
cooling rate of not lower than 1°C/sec. to a temperature
of not higher than 550°C.

20   8.   A method for producing a high-strength steel
pipe excellent in low temperature toughness and weld
heat-affected zone toughness according to claim 7,
characterized by: cold-forming a cooled steel plate into
a pipe; and thereafter applying seam welding to the
25   abutted portion thereof.

    9.   A high-strength steel pipe excellent in low
temperature toughness and weld heat-affected zone
toughness, characterized by, in the pipe having a seam-
welded portion: the base steel containing, in mass,

30   C: 0.02 to 0.1%,
    Si: not more than 0.8%,
    Mn: 1.5 to 2.5%,
    P: not more than 0.015%,
    S: not more than 0.003%,
35   Ni: 0.01 to 2%,
    Mo: 0.2 to 0.8%,
    Nb: less than 0.010%,

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Ti: not more than 0.03%,

Al: not more than 0.1%, and

N: not more than 0.008%,

with the balance consisting of Fe and unavoidable

5   impurities; the P value defined by the following

expression being in the range from 1.9 to 4.0; and the

microstructure being mainly composed of martensite and

bainite:

$$P = 2.7C + 0.4Si + Mn + 0.8Cr + 0.45(Ni + Cu) + 2V +$$

10    $$Mo - 0.5.$$

10. A high-strength steel pipe excellent in low

temperature toughness and weld heat-affected zone

toughness, characterized by, in the pipe having a seam-

welded portion, the base steel containing, in mass,

15   C: 0.02 to 0.10%,

Si: not more than 0.8%,

Mn: 1.5 to 2.5%,

P: not more than 0.015%,

S: not more than 0.003%,

20   Ni: 0.01 to 2%,

Mo: 0.1 to 0.8%,

Nb: less than 0.010%,

Ti: not more than 0.030%,

B: 0.0003 to 0.003%,

25   Al: not more than 0.1%, and

N: not more than 0.008%, so as to satisfy the

expression $Ti - 3.4N \geq 0$,

with the balance consisting of Fe and unavoidable

impurities; the P value defined by the following

30   expression being in the range from 2.5 to 4.0; and the

microstructure being mainly composed of martensite and

bainite:

$$P = 2.7C + 0.4Si + Mn + 0.8Cr + 0.45(Ni + Cu) + 2V +$$

$$1.5Mo.$$

35   11. A high-strength steel pipe excellent in low

temperature toughness and weld heat-affected zone

toughness according to claim 9 or 10, characterized by

further containing, in mass, one or more of

        V: 0.001 to 0.3%,

        Cu: 0.01 to 1%,

        Cr: 0.01 to 1%,

5   Ca: 0.0001 to 0.01%,

        REM: 0.0001 to 0.02%, and

        Mg: 0.0001 to 0.006%.

        12.   A high-strength steel pipe excellent in low temperature toughness and weld heat-affected zone

10 toughness according to any one of claims 9 to 11, characterized by the average diameter of the austenite grains in the steel pipe being not larger than 10 μm.

        13.   A high-strength steel pipe excellent in low temperature toughness and weld heat-affected zone

15 toughness, characterized by, in the pipe having a seam-welded portion: the base steel containing, in mass,

        C: 0.02 to less than 0.05%,

        Si: not more than 0.8%,

        Mn: 1.5 to 2.5%,

20   P: not more than 0.015%,

        S: not more than 0.001%,

        Ni: 0.01 to 2%,

        Mo: 0.1 to 0.8%,

        Nb: less than 0.010%,

25   Ti: not more than 0.030%,

        B: 0.0003 to 0.003%,

        Al: not more than 0.1%, and

        N: not more than 0.008%, so as to satisfy the expression Ti − 3.4N ≧ 0, and further one or more of

30   V: 0.001 to 0.3%,

        Cu: 0.01 to 1%, and

        Cr: 0.01 to 1%,

with the balance consisting of Fe and unavoidable impurities; the P value defined by the following

35 expression being in the range from 2.5 to 4.0; the microstructure being mainly composed of martensite and

bainite; and the average diameter of the austenite grains being not larger than 10 μm:

$$P = 2.7C + 0.4Si + Mn + 0.8Cr + 0.45(Ni + Cu) + 2V + 1.5Mo.$$

5   14.   A high-strength steel pipe excellent in low temperature toughness and weld heat-affected zone toughness, characterized by, in the pipe having a seam-welded portion, the base steel containing, in mass,

C: 0.02 to less than 0.05%,

10   Si: not more than 0.8%,

Mn: 1.5 to 2.5%,

P: not more than 0.015%,

S: not more than 0.003%,

Ni: 0.01 to 2%,

15   Mo: 0.1 to 0.8%,

Nb: less than 0.010%,

Ti: not more than 0.030%,

B: 0.0003 to 0.003%,

Al: not more than 0.1%, and

20   N: not more than 0.008%, so as to satisfy the expression $Ti - 3.4N \geqq 0$, and further one or more of

V: 0.001 to 0.3%,

Cu: 0.01 to 1%,

Cr: 0.01 to 1%, and

25   Ca: 0.0001 to 0.01%,

with the balance consisting of Fe and unavoidable impurities; the P value defined by the following expression being in the range from 2.5 to 4.0; the microstructure being mainly composed of martensite and

30   bainite; and the average diameter of the austenite grains being not larger than 10 μm:

$$P = 2.7C + 0.4Si + Mn + 0.8Cr + 0.45(Ni + Cu) + 2V + 1.5Mo.$$

      15.   A method for producing a high-strength steel

35   pipe excellent in low temperature toughness and weld heat-affected zone toughness, characterized by: reheating

the casting containing components according to any one of claims 9 to 14, to a temperature of not lower than the Ac_3 point; hot rolling it; thereafter cooling the resulting steel sheet at a cooling rate of not lower than

5   1°C/sec. to a temperature of not higher than 550°C; cold-forming the cooled steel plate into a pipe; then applying submerged arc welding to the abutted portion from the outer and inner sides thereof; and thereafter subjecting the steel pipe to pipe expansion.

10   16. A method for producing a high-strength steel pipe excellent in low temperature toughness and weld heat-affected zone toughness according to claim 15, characterized by heating the seam-welded portion of the steel pipe to 300°C to 500°C before pipe expansion.

15   17. A method for producing a high-strength steel pipe excellent in low temperature toughness and weld heat-affected zone toughness according to claim 15, characterized by heating the seam-welded portion of the steel pipe to 300°C to 500°C after pipe expansion.

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HIGH-STRENGTH STEEL EXCELLENT IN LOW TEMPERATURE
TOUGHNESS AND TOUGHNESS AT WELD HEAT-AFFECTED ZONE,
METHOD FOR PRODUCING THE SAME,
AND METHOD FOR PRODUCING HIGH-STRENGTH STEEL PIPE

5

ABSTRACT OF THE DISCLOSURE

10

The present invention provides an ultra-high-
strength steel pipe excellent in weldability on site and
a method for producing the steel pipe by improving the
reliability of the low temperature toughness of a steel
15 to which elements to enhance hardenability are added for
furthering high-strengthening and also improving
toughness at a weld heat affected zone subjected to
double or more layer welding and, in the method, the
steel is made to consist of a structure composed of
20 bainite and/or martensite by containing prescribed
amounts of C, Si, Mn, P, S, Ni, Mo, Nb, Ti, Al and N,
and, as occasion demands, one or more of B, V, Cu, Cr,
Ca, REM and Mg, and regulating C, Si, Mn, Cr, Ni, Cu, V
and Mo, those being elements to enhance hardenability, by
25 a specific relational expression. The diameter of prior
austenite grains may be regulated in a prescribed range.
The method includes the steps of heating a casting to a
temperature not lower than the Ac_3 point, hot rolling it,
and thereafter cooling the resulting hot-rolled steel
30 plate at a prescribed cooling rate.

Fig.1